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Eva MAZANCOVÁ[*], Stanislav RUSZ[**], Zdeňka RUCKÁ[***], Karel MAZANEC[****]

ACICULAR FERRITE AND BAINITE MICROFRACTOGRAPHIC RESPONSE

MIKROFRAKTOGRAFICKÁ ODEZVA ACIKULÁRNÍHO FERITU A BAINITU

Abstract

The acicular ferrite (AF) and bainite (B) formation is practically connected with the same temperature range. The plates (laths) of the both austenite transformation products show displacive mechanism character. The differences are in nucleation sites. At the non-metallic inclusions or precipitates, the intragranular nucleated AF leads to the microstructure refinement consisting of interlocked plates (laths). The B is formed intergranularly. In impact toughness specimens (CVN-60) the fractal analysis of fracture surfaces was performed. In the AF the determined values of fractal dimenses are higher (cleavage fracture) than in the B microstructure. In the zone of ductile fracture, determined difference of fractal dimension is small.

Abstrakt

Vznik acikulárního feritu (AF) i bainitu (B) je spojen prakticky se stejným teplotním intervalem. Desky (laťky) obou produktů austenitické transformace vykazují displacivní charakter. Rozdíly jsou v pozicích nukleace. Intragranulární AF vzniká na nekovových částicích, vykazuje propletené desky (laťky), což vede ke zjemnění mikrostruktury. Bainit je tvořen intergranulárně. Fraktální analýzy byly provedeny na lomových plochách zkoušek vrubové houževnatosti realizované při -60°C. V případě AF jsou stanovené hodnoty fraktální dimenze vyšší (větší podíl tvárných hřebenů) než v mikrostruktuře s B. Rozdíl fraktálních dimenzí ve tvárném lomu (v blízkosti čela vrubu) je malý.

1 INTRODUCTION

Matrix with a microstructure mainly consisting of acicular ferrite (AF) provides an optimum combination of increased strength and beneficial toughness level. This behaviour is given by grain size refinement and by high angle grain boundaries [1-3]. The morphology of the AF differs from that of upper bainite (B), because the AF nucleates intragranularly at non-metallic inclusions or precipitates within large austenite grains [4]. The upper B initiates at austenite grain boundaries and forms parallel laths (plates) with the same crystallographic orientation. The AF produces a microstructure in which the plates (laths) grow in an interlocked arrangement. The AF microstructure offers a fine effective grain size, high angle boundaries and a high dislocation density, resulting in strength and toughness improvement including a favourable hydrogen response [1-3, 5]. At the non-metallic inclusions, together with the sympathetic nucleation, the AF formation process leads to special arrangement of ferritic laths and to fine-grained interlocking microstructure formation. This arrangement of

[*] Prof., Ing. , CSc., Dep. of Materials Engineering., Fac. of Metallurgy and Materials. Engineering., VŠB-TU of Ostrava, Tř. 17. listopadu, Ostrava, tel. (+420) 59 699 4409, eva.mazancova@vsb.cz
[**] Prof., Ing. , CSc., Dep. of Mechanical Technology, Fac. of Mechanical Engineering, VŠB-TU of Ostrava, Tř. 17. listopadu, Ostrava, tel. (+420) 59 699 3315, stanislav.rusz@vsb.cz
[***] Ing., Dep. of Materials Engineering, Fac. of Metallurgy and Materials Engineering., VŠB-TU of Ostrava, Tř. 17. listopadu, Ostrava, tel. (+420) 59 699 4199, zdenka.rucka.st@vsb.cz
[****] Prof., Ing. , DrSc., Dep. of Materials Engineering, Fac. of Metallurgy and Materials Engineering., VŠB-TU of Ostrava, Tř. 17. listopadu, Ostrava, tel. (+420) 59 699 4409, eva.mazancova@vsb.cz

non-parallel laths results in the toughness improvement as, unlike the B, the propagating crack encounters laths in different crystallographic orientations [6, 7].

The aim of presented work is to investigate a modification of the AF microstructure behaviour in comparison with the B as it follows from the microstructure properties of this austenite decomposition product.

2 EXPERIMENTAL APROACH

The work is devoted to the fracture surface evaluation of impact toughness specimens (CVN) using the fractal analysis with a preferential view to the study of steel behaviour in the area of cleavage fraction formation. For the investigation, material of the C-Mn steel rolled into thick sheet (19 mm in thickness) was used. The conventional rolling process (finished at 691°C followed with a cooling on the air) led to final ferrite microstructure with bainite regions in strips. The chemical composition was following (in wt. %): 0.079C; 1.66Mn; 0.35Si; 0.015P; 0.004S; 0.02Cu; 0.02Ni; 0.04Cr; 0.052V; 0.037Nb; 0.008Ti; 0.027Al_t. The calculated critical temperatures equalled to 874°C, 716°C, and 731°C (in sequence the Ac_3, Ac_1 and Ar_3). In laboratory conditions a simulation of finishing rolling temperature above the Ar_3 one was carried out. Within the extent of 780-500°C the accelerated cooling process (ACC with cooling rate of 18.9°C.s^{-1}) led to the formation of the 67 % AF in basic matrix. Details were described formely [4]. The Figs. 1 and 2 represent micrographs of the steel with the B and the AF.

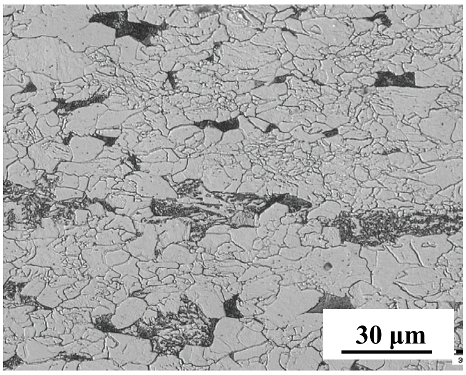

Fig. 1 Micrograph of matrix with the B

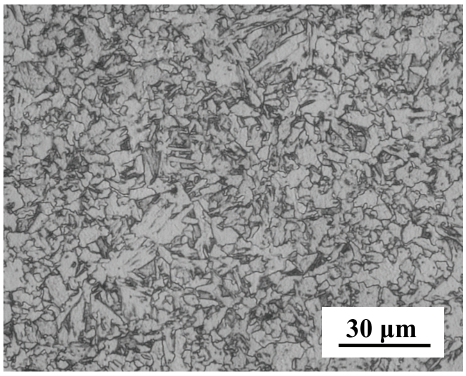

Fig. 2 Micrograph of matrix with the AF

Tab. 1 The reached mechanical properties

Final state	YS	TS	A_5	HV30	$CVN_{(-60)}$
	[MPa]		[%]	[-]	[J]
B	521	632	25.9	228	56
AF	535	669	26.9	239	115

The Tab. 1 summarizes results of the basic mechanical properties. Tensile specimens with a section of 6 mm diameter and gauge length of 30 mm were machined in the rolling direction and the tensile properties were determined using an MTS (100 kN) machine. The hardness (HV30) was also measssured. Conventional V-notch (2 mm in depth) Charpy specimens were perpendicularly machined to rolling direction and tested at -60°C by use of the PSW 300AF machine with maximal capacity of 300 J. The light and electron microscopy was an integral part of metallographic and microfractographic solution. The Figs. 3 and 4 demonstrate studied fracture surfaces. In either case, the fracture surfaces showed cleavage features for the most part, naturally with different morphology of cleavage facets in compared microstructures.

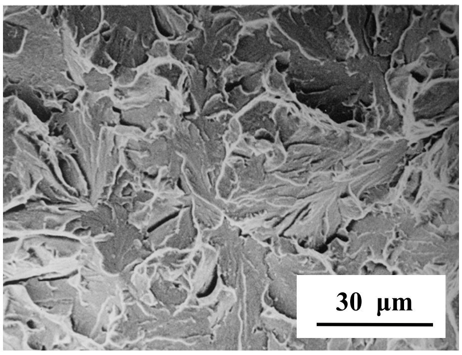

Fig. 3 Fracture surface of the B

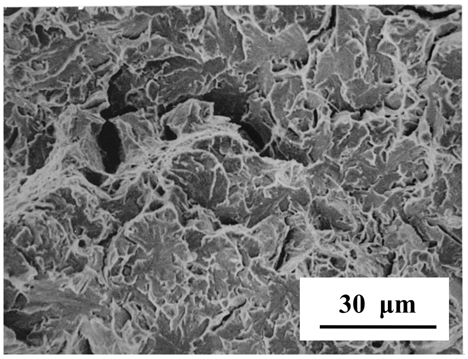

Fig. 4 Fracture surface of the AF

The fractal analysis of fracture profile was carried out in vertical direction. The surface was covered with a protective 50 µm thick Ni-layer. The evaluation was made (electrospark) in 4 vertical sections realized in 0.7, 1.5; 3 and 5 mm distance from the notch-tip of specimen. The aim of this measurement was to obtain the data concerning the fracture behaviour at lower temperature. In strain line the basic distance was chosen of 200 µm. The length of profile lines was evaluated using following magnifications: 200x, 500x and 800x. The evaluation was based, as in former work [8], in implementing the Mandelbrot-Richardson equation [9]:

$$L = L_0 \cdot \varepsilon_i^{1-D} \tag{1}$$

The L is the measured length varying with the ε_i, L_0 represents a constant and the $\varepsilon_i = \eta/L_0$, where η is step-size corresponding to the length of measuring unit. Substituting the ε_i into Eq. (1) we obtain:

$$L = L_0^D \cdot \eta^{1-D} \tag{2}$$

L_0 can be considered as a constant hence the Eq. (2) can be written as:

$$1 - D = d - \log L / \log \eta \tag{3}$$

The slope of log L vs. log η can be used to determine the fractal dimension D. The manual measurements of fractal profile were carried out in graded step size being of 1.5÷30 µm.

3 RESULTS

The dependences of log L on log η are plotted in Figs. 5 and 6 which were determined both in the distance x of 0.7 mm and of 1.5 mm and of 3 mm from crack-tip of impact toughness test piece. In these distances the features of fracture surfaces correspond to the cleavage fracture while on the contrary, the ductile fracture formation (stable crack growth) was found in the distance of 0.7 mm. The summarized results of the D-measurements found in chosen distances x from the notch-tip of specimens are plotted in Fig. 7. This figure demonstrates the changes of the D-values in the matrix with the AF being realized more slowly in dependence on distance from the notch-tip in comparison with the modification of the D-values determined in microstructure with the B. The D-values found in distance of 0.7 mm from the notch-tip differ slightly (1.22 and 1.25 for the AF and for the B). The special behaviour of the matrix with the AF (in case of cleavage fracture formation) results both in higher resistance to crack formation and in its propagation, like also in achievement of the higher impact toughness (CVN) values. The small difference in the D-values as obtained in distances x of 1.5 and 3 or 5 mm shows the cleavage fracture being realized under increased difficulties in the matrix with the AF as it follows from its special morphology arrangement (D=1.21 and the further values are 1.19). In case of the matrix with the B, the attained D-values equal to 1.15÷1.135 in distance x=1.5÷5 mm. It represents by 4.8-5.2 % lower D-values in comparison with the AF matrix as it follows from the Fig. 7.

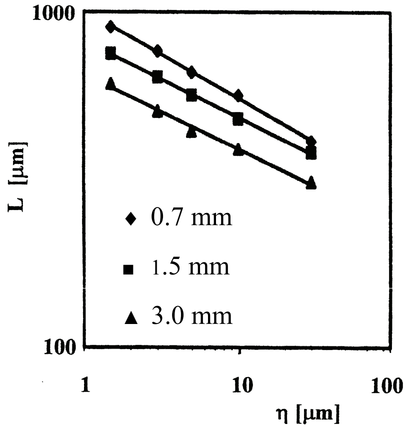

Fig. 5 Dependence of log L vs. log η (the AF)

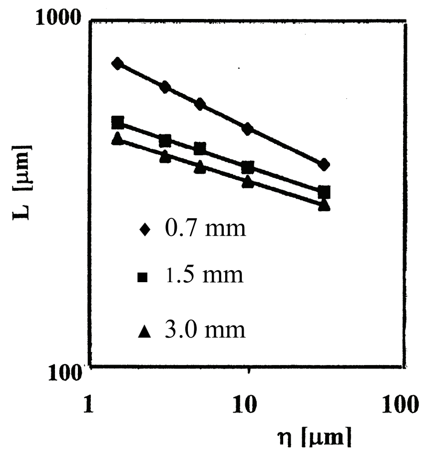

Fig. 6 Dependence of log L vs. log η (the B)

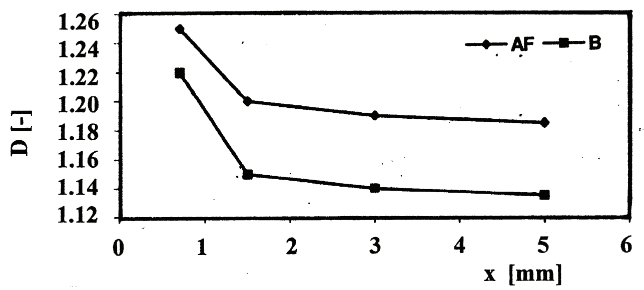

Fig. 7 Fractal dimension (D) vs. distance (x) from the notch-tip

The results demonstrate (to certain extent surprisingly) very near level of the D-values found in the zone of prevailing ductile fracture formation, what is not in an expected relation to CVN-values. In this connection, it is necessary to take into consideration a possibility of certain disproportion of the D-values with the data concerning the resistance level to fracture formation (as was found in some measurements of fractal parameters). On the contrary, as it follows from former evaluations, a great difference in the distribution of plastic deformation under fracture surface can be observed. The grain refinement of the matrix with the AF (Fig. 2) can contribute to the more intensive development of plastic deformation under fracture surface.

The higher level of the determined D-values of specimens consisting of matrix with the AF in a predominant degree evidently reflects the effect of microstructure refinement and morphology arrangement of this phase particles, in particular in a range of the η-lowest values. In this case, the individual deviations are sensitively determined in fracture profile line. The effect of the grain size refinement is partially "getting lost" if the larger step-sizes are applied. This results in higher slope like of (d-log L/log η) dependence, like in higher D-values. On the contrary, a matrix with the B is characterized by easy crack propagation in the volume of the individual B shaves [5]. The mentioned range represents the interfaces having a high angle character what leads to the limiting of the crack propagation. The cracks deviations during their propagation are also less probable because the interaction of growing cracks with carbide precipitates is limited as it follows from the special B morphology characterized by larger free paths for cleavage crack propagation in laths of the B. Even, in connection with above mentioned different morphology of the B and of the AF, the B formation can be held for a special form of degradation process when we take into consideration that the both compared types of the displacive austenite decompositions are realized at the same temperature range. The different morphology of compared phases follows from the difference of nucleation mechanisms.

4 DISCUSSION

The presented work, though gives data about the *D* for matrix with the AF and the B, is devoted to the discussion of morphology influence on the resistance to fracture preferentially. Decisive parameter is the difference in nucleation conditions of considered products of displacive austenite decomposition. The AF is intragranularly initiated at non-metallic inclusions situated in the volume of the austenite matrix. The formed AF particles are "fan-shaped" arranged what in comparison with the B means "a priori" particular complication by the process of crack growth (evidently preferentially by cleavage cracking).

Besides this mechanism, the AF formation process is further accompanied with initiation of secondary ferrite particles nucleated on some primary AF laths (so called sympathetic nucleation). It leads to a formation of chaotic lath arrangement and fine grained interlocking microstructure in which the free paths for crack propagation are very short. This action results in the more frequent cracks deviations during their propagation (the higher D-values found by cleavage fracture formation - Figs. 5 and 7). In case of the B formation an easy crack growth is realized in the volume of individual sheaves. The special behaviour of the AF-laths interfaces can be held for a very important parameter. While the B ferrite interfaces are of low angle type (only the interfaces of sheaves are of high-angle type), in case of the AF, the high angle interfaces are formed between adjoining laths. Especially those latter mentioned interfaces support the crack deviation and represent a certain hindrance form for crack growth.

The above discussed AF microstructure arrangements lead to the higher resistance of cleavage crack propagation and within framework of presented solution to the finding of the higher *D*-values. Concerning the ductile fraction formation, the refinement of the AF laths can play a positive role in detected beneficial response in comparison with the B microstrucutre berhaviour (without taking into consideration the plasticity development under fracture surface).

The above given statements have certain limitation. Primarily, it is morphology refinement of the AF laths and their arrangement. The AF-lath morphology can be modified in dependence on the nucleability properties of non-metallic inclusions [9]. The important requirement is the more elongated AF laths formation. Thereby, more opportunity for edge on face nucleation can be provided and the above mentioned interlocking (interwoven) morphology arrangement can be promoted. Simultaneously, the described process makes possible to attain a higher level of microstructure "filling" with the AF laths resulting in a lower volume fraction of microphase (having e.g. a higher C-content, inclusive of the possible M/A constituent formation and/or ferrite/cementite aggregates [7]). The microphase means any transformation product in isolated regions left between impinged AF-laths due to special microstructure arrangement. It is advisable to hold the volume fraction of microphase under 10 %, approximately, representing an optimum condition. This microphase formation in a higher volume fraction is accompanied with a localised appearance of larger facets at fracture surface (sometimes can be observed in evaluated vertical fracture section). The facets size is usually 15÷25 μm on average and corresponds to the facets found in the B microstructure. The facets measured in the AF microstructure have 3 and/or 5 times smaller size [5]. In this connection, it is useful to remark that the morphology arrangement consisting of shorter AF laths limits the secondary ferritic particles formation by way of symphatetic nucleation. The beneficial effect ascribed to the interlocking effect is less probable in this case. The process of prolonged lath formation is sufficient in contrast to the short laths formation being ascribed to the lower driving force for the lath growth. The potential surfaces for sympathetic nucleation are not at disposal.

The discussed results show that the fractal analysis of numerical data concerning the level of separate fractal dimensions, can effectively contribute to the accomplishing evaluation (special aspects) of microstructure characteristics. This solution is preferentially devoted to the knowledge elaboration concerning the mutual relationship between mechanical-metallurgy properties and microstructure including the requirements on the non-metallic nucleability and its influence on the products of the austenite decomposition. The presented study includes the physical metallurgy analysis of the

AF and the B microstructure parameters and their influences on the microfractography behaviour. The applied fractal analysis is chosen as an additional method contributing to the achievement of optimum microstructure properties [1, 7].

5 CONCLUSIONS

The work summarises the results obtained by complex evaluation of mutual relationship between microstructure features of the C-Mn steel. Its microstructure is consisting either of the AF or of the B microstructure. The different behaviour of these microstrucutres was elucidated on the basis of their fracture response comparison. This fracture behaviour is given by the different morphology parameters of these austenite decomposition products through the temperature nucleation region and the displacive mechanism of the both compared microstructures. The controlling parameter is the difference in nucleation sites (inter- and/or intragranular nucleation in the austenite matrix). The causes influencing the formation of different fracture response of the AF and of the B are elucidated on the basis of the microstructure analysis and metallurgy parameters realized by cleavage crack propagation.

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